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Magnesium.

THE production and use of magnesium and its alloys have extended so rapidly during recent years that there may be a tendency to forget the long period of preparation which made this development possible. It is not easy to realise that the centenary of the first electrolytic separation of magnesium from its chloride will soon be upon us. Although sought for by Davy and isolated by him in 1808, the relatively pure metal was first won in amounts large enough to enable a study of its properties to be made by Bussy in 1830 and by Bunsen in 1852. Following in their footsteps others have studied magnesium and utilised its remarkable properties in countless ways. But it was during and after the Great War that magnesium was granted a place among the industrial metals and that its alloys began to attract attention on a wide scale. The labours involved in attaining to the present position of the industrial production and use of magnesium have been very great, and much credit is due in this connection to the Griesheim Elektron Company in Germany, and to the Dow Chemical Company of America.

There is little doubt that these industrial developments stimulated research on magnesium alloys. The practical value of such research, however, begins to emerge only when the results are absorbed and found to be capable of utilisation by the technical metallurgists and engineers on whom the successful progress of our industries so largely depends.

The autumn meeting of the Institute of Metals at Newcastle has shown that research and experimental work on magnesium is being given the high place it deserves in the proceedings of the Institute. Four of the papers were concerned with this metal and its alloys, thereby providing opportunity for discussion, for exchange of ideas, and for the sharing of new information. The remarks of Sir Charles Craven, one of the foremost leaders of engineering industries of the North, to the Institute gave strong assurance of the eagerness of the industries to make full use of the latest scientific advances.

Each new day sees greater use of light metals and alloys in engineering and other industries. In some applications the properties of magnesium and magnesium-rich alloys place them in a position of advantage over aluminium and aluminium alloys. The most outstanding characteristic is their lightness. As examples one might mention that magnesium alloy castings in common use have only two-thirds the weight of aluminium alloy castings of the usual types, and only one-quarter of the weight of iron castings. In all cases where burdens have to be carried or matter transported the attractiveness of saving in weight is inescapable. Particularly in the case of parts of machines which have to be moved at high speeds is the matter of weight reduction of importance since the possibility of reduced bearing pressures, inertia loads, centrifugal loads, &c., presents itself immediately.

The alloys can be extruded and rolled with advantage to their mechanical properties. Forgings, stampings, and pressings are being produced in wide variety. At present the capacity of the alloys to withstand cold forming operations is very limited, but severe profiling and forming is possible at elevated temperatures. As regards the limited capacity of the alloys to withstand cold working, the position cannot, in view of the comparative youth of the alloys, be regarded as settled. Here is a field in which metallurgical research may well prove its worth. The machining properties of magnesium alloys constitute another feature of high importance. Machining is possible at higher speeds and more easily than with any other industrial metal. In many cases this quality has more than offset the higher cost of the raw material as compared with cast iron and other metals.

The satisfactory resistance to corrosion of magnesium alloys under atmospheric exposure and normal conditions of use has now been established by ample demonstrations and experiences extending over many years. The improvement in corrosion properties with improved methods of manufacture and handling has been a surprise to many a metallurgist. The fact is that we have in magnesium alloys a group of materials which bear little resemblance to their forebears of ten years ago, but compare favourably with the much-used materials of engineering construction in the matter of corrosion resistance. The alloys are, however, liable to corrode on prolonged exposure to humid conditions or marine exposure. For such conditions, protective treatment is necessary, and is available in a variety of forms. Protective surface films produced by simple chemical immersion treatments, supplemented by suitable enamels or varnishes, are giving satisfactory protection under exacting conditions of use. The resistance of the alloys to alkalis and to many organic chemicals is remarkable and likely to prove of value.

Most of us have seen thin magnesium ribbon burn, and the question of fire risk naturally arises in the mind. In order to ignite magnesium it is necessary to raise the temperature to the melting point, or, in the case of alloys, to the temperature at which melting begins. For continuous burning to occur the conditions have to be such as to allow very ready access of air to the metal. Filings, turnings, and thin ribbon can readily be ignited, but larger masses do not burn except when the surroundings are at a high temperature or when other substances are burning about it in large quantity, *i.e.*, under conditions which would cause the collapse of parts made in any other light alloy. Fires of turnings

started accidentally or by the use of unsuitable conditions of machining do not spread rapidly, and are readily quenched by simple methods. It is, however, very necessary to employ methods suitable for magnesium since many common methods of fire extinction fail, in this instance, to have the desired effect. The way in which magnesium alloys are now melted in open crucibles in ordinary gas and oil-fired furnaces in our foundries, and the welding of magnesium alloy sheets and sections, as now practised with the oxy-acetylene torch, have done much to remove doubts and suspicions from the minds of many potential users. Besides the question of fire risks there is another workshop problem which is exercising the attention of magnesium workers. In shops where fine machining is done it has been found that owing to its extreme lightness fine magnesium dust remains suspended in the air for a long time and may be inhaled by the workpeople with danger to the health. The problem is a new one in workshop hygiene and may call for special measures.

In attempting to visualise the scope of future progress it may be helpful to look back over the past few years. That magnesium-rich alloys have made remarkable and steady progress is beyond dispute. That the progress is continuing unabated at the present time is equally clear. Looking to the future there appears no room for doubt of the high position these alloys must attain.

Metal Melting.

OF all the various processes involved in the manufacture of metallurgical products, melting is one of the most fundamental and important, for it is at this stage that the quality of the material is largely determined. In the production of pure metals and alloys as castings or in fabricated forms, such as sheet, plate, bar, rod, tube, and wire, the process of melting, by which the basis forms, such as ingot, cake, slab, billet, shell, &c., are produced, is to be differentiated from the yet earlier operation of reduction of the metal from its ores and concentrates which also usually involves melting. In considering melting as distinct from smelting processes of extraction, Dr. H. W. Brownsdon, in the Autumn Lecture to the Institute of Metals at Newcastle-on-Tyne, discussed generally the various factors involved, which vitally affect the quality of the resulting material. Departing from the usual procedure, the lecture was followed by a discussion in which the speakers for the most part each dealt with some specific aspect of the subject. The influence of gases, the use of fluxes, deoxidants, and other additions, contamination by impurities from different sources, such as refractories and fuels, were amongst the matters discussed. The subject of gases in metals has at different times received quite a lot of attention, but it is very evident that much remains to be found out. On the effect of fluxes, deoxidants, and other special additions, very little information indeed is available. In certain applications, treatment of molten metal by gases has been most successful in promoting increased soundness and improved properties, and this is clearly a line of investigation from which further valuable results may be expected. It has, in fact, been found possible in commercial practice to improve quality by gas treatment, careful temperature control, use of suitable deoxidants, controlled rates of pouring and the use of carefully selected materials in the heats, along lines indicated by researches which have already been carried out, but much more work needs to be done. In spite of the development which may be expected to take place in powder metallurgy where the melting operation is eliminated, and in operations such as the Hazlett process, where casting in the ordinary sense of the word is not carried out, the usual processes of melting and casting will doubtless long remain metallurgical operations of primary importance. Not only are fundamental data required regarding gases in metals, equilibria between fluxes, the contents of the bath, and deoxidation processes, but also more information is needed of a practical kind regarding the most satisfactory melting procedures for different metals and alloys. Discussions of the kind which have taken place are undoubtedly most helpful in effecting a dissemination of useful practical knowledge and in focusing on urgent problems the simultaneous attention of research workers and those engaged in actual production. In addition to the several factors involved in melting, there are also many factors associated with the operation of casting, including rate and temperature of pouring, type of mould, and mould materials temperatures and dressings, which are of considerable importance. When one reflects on the large number of possible influences which may affect the quality of the product at this stage, one cannot help the thought that if, in spite of all these metallurgical risks and dangers, it is possible with some materials at least to achieve a moderate degree of success, how great an improvement might reasonably be hoped for if more were known of these various factors and how they affect different metals and alloys. Although the difficulties encountered in melting, and the production of cast material of satisfactory quality are only too well known to those actively engaged in production, the lecture and the ensuing discussion have served the valuable purpose of itemising very clearly the several factors involved. These are so numerous that it is easy for many of them to be obscured. The debate has also served to draw the attention of the scientific investigator to matters and problems of technical interest and vital commercial significance which call for attention and await solution. In so doing it is hoped that it will stimulate further work leading to a more perfect understanding and more effective control of the influences involved in melting and casting which affect the quality and properties of metals. Important though it is, there are other considerations in metal melting besides the quality of the material produced, and chief amongst these are the economics of the process. These are determined in some measure by the types of furnaces and fuels used and by shop practice. In the choice of the former the producer is often limited by local conditions; whilst whether, for example, the melting is continuous or intermittent is often largely determined by local and transient circumstances. There are, however, factors effecting the economics of melting processes which are capable of control at least to some extent. One of these is the loss of metal which occurs. While this may be expected to vary with the composition of the alloy and the type of furnace in which it is melted, that other factors affecting it can be controlled with very beneficial results has been clearly shown in the paper presented to the same meeting by Dr. M. Cook, who gave the results of a careful analysis of the sources of loss in the melting of copper alloys, both in coke-fired crucible furnaces and Ajax Wyatt induction furnaces, and of various expedients tried for reducing the metal

loss. With these materials the loss is largely determined by the extent to which oxidation can be prevented and the degree of care exercised in handling during melting and casting. The very ready and informative discussion which followed indicated how general was the interest in the application of scientific methods of inquiry and control to the economic side of melting metal, and it is to be hoped that more papers of a similar practical kind will be forthcoming, for there can be no doubt that in the discussion which such papers tend to promote the interchange of information and experience plays no small part in stimulating technical progress.

Case-Hardening Alloy Steels.

The influence of a variety of alloy additions on the case-hardening properties of steel containing 0·15 to 0·25 per cent. of carbon has been described by E. Houdremont and H. Schrader in a paper on "The Influence of Alloy Elements on the Behaviour of Steels in Case Hardening."*

The alloy steels were made in crucibles, cast into ingots 100 mm. square, rolled to 45 mm. diameter, and carburised as test pieces, 40 mm. diameter by 100 mm. long, in a case-hardening mixture of 60 per cent. wood charcoal and 40 per cent. barium carbonate at 830–850, 900–920, and 980–1000 deg. Cent. for ten, thirty, and sixty hours. The maximum carbon content of the case, depth of penetration of carbon, depth of hardening, structure, and sensitivity to overheating were all determined. The results are shown in a series of curves in the original paper, and the effect of the alloy additions on the structure are illustrated by some fine composite photomicrographs of sections of the case-hardened bars.

Maximum Carbon Content of the Case.

Houdremont and Schrader found no connection between the degree of surface carburisation and the extent of the gamma region as affected by the added element. Elements which enlarge the gamma region and those which restrict it often have the same effect on the behaviour of the steel in case hardening. When, however, the elements are arranged according to their tendency to form special carbides they fall into a series, with vanadium at one end and copper or silicon at the other, which is characterised by a gradually decreasing effect on the maximum carbon in the case. Thus, after sixty hours' carburising at 900 deg. Cent. of steels containing 3 per cent. of the added element, the following maximum carbon contents were obtained in the case, the plain carbon steel giving 1·4 per cent.:—

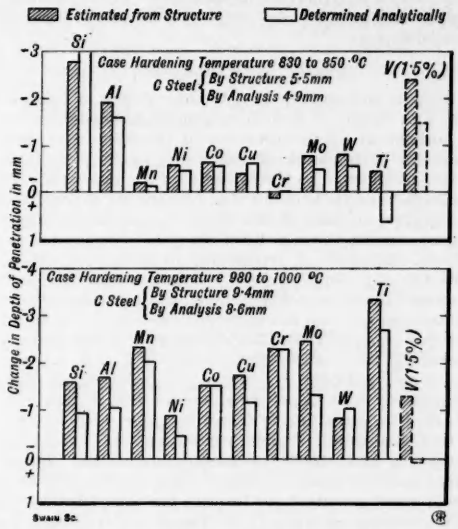
Si.	Cu.	Ni.	Mn.	Al.	W.	Cr. Mo. Ti.
0·9 ..	1·2 ..	1·2 ..	1·25 ..	1·35 ..	1·6 ..	1·75

All special carbide-forming elements, such as chromium, molybdenum, tungsten, vanadium, and titanium, produced an increase in maximum carbon content, whereas additions of silicon, aluminium, nickel, cobalt, and copper, as well as the impurities, sulphur and phosphorus, resulted in a lowering.

In general, the maximum carbon content of the case increased with temperature and time of carburising, though there were some irregularities, and in particular the maximum carbon contents attained in the steels giving the highest and the lowest figures, viz., the 3 per cent. titanium and 3 per cent. silicon steels, were not much affected by varying the carburising conditions from ten hours at 830 deg. Cent. to sixty hours at 1000 deg. Cent.

Depth of Penetration of Carbon.

The depth of penetration was lowered by the majority of the alloy elements studied. For steels containing 3 per cent. of the alloy element the change in depth of penetration compared with that of the carbon steel after sixty hours' carburising at about 850 deg. and 1000 deg. Cent. is shown in the diagram. At the lower temperature silicon and, to a less extent,



Change in Depth of Penetration of Carbon Produced by an Alloy Addition of 3 per cent. in a 60 Hour Case-Hardening Treatment.

aluminium hinder the penetration of carbon. Vanadium is also detrimental to carbon penetration; manganese and chromium have little influence and the remaining elements produce only an unimportant decrease in the depth of penetration. The depth of carbon penetration was estimated from the structure and also by chemical analysis. There was a discrepancy between the results of these two methods as applied to some of the steels, *e.g.*, the vanadium, molybdenum, and tungsten steels, and particularly titanium steel, which was judged by structure to diminish, but by analytical methods to increase, the depth of penetration. This is due to the fact that small aggregated particles of carbide are overlooked in the structure, and the chemical method, of course, yields the more correct results. At the higher temperature of cementation the effect of silicon, aluminium, and vanadium in reducing depth of penetration is not so great as at lower temperatures, but the effect of the other elements is in general increased. The influence of nickel and of tungsten in restricting the depth of penetration is less than that of the other elements, but at 1000 deg. Cent. manganese, chromium, molybdenum, and titanium have a powerful effect in reducing carbon penetration.

Diffusion of Carbon.

The diffusibility of carbon in the alloy steels can only be estimated from these experiments by taking

* *Archiv für das Eisenhüttenwesen* (April, 1935), 1934–35, 8, 445–459.

into account both degree of surface carburisation and depth of penetration. When the maximum carbon in the case is high, as with chromium, molybdenum, and tungsten, a decrease or even a constant value for depth of carbon penetration indicates reduced diffusibility. With a lower surface carbon content a decrease in depth of penetration can be explained exclusively as a result of smaller concentration gradient; this applies to silicon and aluminium, which thus do not appear to reduce diffusion. When there is only a very small decrease in carbon penetration, in spite of a lower carbon content at the surface, as with nickel, cobalt, and copper, the indication is that these elements tend to increase diffusibility. For cobalt and copper this applies mainly to the lower carburising temperatures.

DEPTH OF HARDENING.

Several factors help to determine depth of hardening, *e.g.*, depth of carbon penetration, hardenability as influenced by the presence of the alloy elements, and the withdrawal of hardening carbon to form stable carbides. To throw light on the interaction of these factors Houdremont and Schrader determined the carbon content at the limit of the case-hardened region nearest the core after suitable heat treatment, which consisted of hardening in water or oil at 760–850 deg. Cent. In the specimens case-hardened for sixty hours at 900 deg. Cent. the hardened region ended at a position having a carbon content of about 0·5 per cent. in the following steels:—Carbon steels including those with phosphorus, 0·2 and 0·6 per cent. and sulphur 0·2 per cent.; and steels with aluminium up to 3 per cent., cobalt 1·5, copper 1·5, silicon 1·5, vanadium 0·8, and tungsten 1·5 per cent. Steels containing higher percentages of copper or tungsten, and steels containing chromium, manganese, molybdenum, or nickel hardened fully at a lower carbon content (0·2 to 0·4 per cent.) as a result of the increased capacity for hardening conferred by these elements. In steel with 1·5 per cent. of vanadium and in titanium steels the carbon content at the margin of the hardened region rose to 0·8 and 1·1 per cent. respectively owing to the formation of special carbides which are dissolved but with difficulty and which do not participate in the hardening.

Thus in several alloy steels like nickel, chromium, or molybdenum steels the lower depth of penetration of the carbon is counterbalanced by a greater capacity for hardening, while a lower penetration of carbon combined with a smaller hardening capacity considerably reduces the depth of hardening in silicon and aluminium steels. The depth of case in steels containing vanadium and titanium was small on account of the presence of carbides unaffected by the hardening treatment applied.

MICROSTRUCTURE.

In steels containing the carbide-forming elements such as chromium, molybdenum, tungsten, and titanium, there is an accumulation of free carbide at the surface. The presence of silicon, aluminium, and nickel tends to promote the decomposition of cementite with the formation of graphite. Grain growth is usually restricted by the presence of manganese, nickel, cobalt, copper, or silicon, and particularly by the carbide-forming elements molybdenum, tungsten, vanadium, and titanium. Steels high in sulphur and phosphorus behave similarly to carbon steels, and the steels containing chromium and aluminium are even more sensitive to overheating than the plain carbon case-hardening steels.

In the ensuing discussion the grain growth found in steels of the higher chromium contents was referred to, and lower case-hardening temperatures, possibly with the use of liquid cementing media, were advised to counteract it. The authors' work was held to show that the frequently expressed opinion that dissolved oxygen made steel sensitive to overheating was incorrect. Bessemer steel in which oxygen is not low is, it was stated, more insensitive to overheating than other types, whereas crucible steel with low oxygen is comparatively susceptible. Vanadium also increased sensitivity to overheating. According to Houdremont it is not so strong a deoxidiser as has frequently been stated; it passed less readily into the slag than chromium, and on remelting vanadium losses were smaller than chromium losses. This was questioned by Rapatz, who held that in the presence of amounts of the order of 0·2 per cent. the opposite was true.

The question was raised as to why a comparatively low maximum manganese content was usually fixed in specifications for case-hardening steels, and it was held to be a relic of the time when temperature control of heat treatment was less exact. According to Houdremont, there were no grounds for this restriction of manganese, and, moreover, Krupps had formerly made a case-hardening steel with 1·2 to 1·5 per cent. of manganese which had proved entirely suitable for purposes for which a great depth of hardening was required.

The Diffusion of Gases Through Metals.

THE "Proceedings" of the Royal Society for May 1st, 1935, contains an important paper on the above subject by Dr. C. J. Smithells and Mr. C. E. Ransley, both of the research laboratories of the General Electric Company. After reviewing a considerable amount of previous work* on the subject, they describe their experimental procedure. Determinations were made on the rates of diffusion of a number of gases, hydrogen, nitrogen, oxygen, &c., through various metal tubes. Nickel tubes, prepared by electrodeposition, followed by drawing and annealing, had the following dimensions:—

	cm.
External diameter	0·2
Wall thickness	0·01
Heated length	4·2

Copper tubes were drawn from oxygen-free conductivity copper and were heated in hydrogen to 600 deg. Cent. for some hours to remove impurities.

	cm.
External diameter	0·12
Wall thickness	0·01
Heated length	11·0

Molybdenum tubes were made from sheet electrically welded.

	cm.
External diameter	0·1
Wall thickness	0·0075
Heated length	3·8

Iron tubes were obtained by decarburising mild steel tubes in wet hydrogen at 870 deg. Cent.

	cm.
External diameter	0·2
Wall thickness	0·02
Heated length	5·5

This material possessed a fine-grained structure, about 100 grains to the square millimetre.

* This work was reviewed by the writer in THE METALLURGIST, 1930, p. 24.

The tube was heated by passing a current through the tube itself and the end effect due to cooling was avoided by reducing the diameter of the leads in contact with the tube so that they were heated to the same temperature as the tube. The open end of the tube was silver soldered to a platinum thimble which was sealed into a glass tube. The latter was sealed into a hard glass diffusion bulb in which gas could be maintained at any particular pressure. The interior of the metal tube was kept continuously evacuated, any gas diffusing through being collected, measured, and analysed. The temperature of the tube was measured by a chromel-alumel thermo-couple which could be moved along the former by means of a magnetic device. The variation of temperature between the electrical leads never exceeded 10 deg. at the highest temperatures.

Richardson, Nicol, and Parnell† deduced an

TABLE I.—*Effect of Temperature on Rate of Diffusion.*

Pressure, mm.	Temp. ° K.	Rate (D).	Pressure, mm.	Temp. ° K.	Rate (D).
		Hydrogen-molybdenum.			
95·4	823	$0·7\times10^{-6}$	95·4	1298	$40·4\times10^{-6}$
	1023	5·1		1353	52·5
	1133	9·0		1473	96·0
	1188	17·3		1523	117·0
	1228	21·6		1623	182·0
	1263	30·0		1673	204·0
4·67	823	$0·12\times10^{-6}$	4·67	1263	$4·56\times10^{-6}$
	1023	1·02		1298	6·42
	1123	1·70		1353	10·5
	1188	2·54		1473	18·2
	1228	3·77		1673	39·8
0·08	1023	$0·12\times10^{-6}$	0·08	1298	$0·82\times10^{-6}$
	1133	0·21		1353	1·46
	1173	0·29		1473	2·26
	1228	0·51		1673	3·90
		Nitrogen-molybdenum.			
130	1373	$0·78\times10^{-7}$	130	1723	$16·5\times10^{-7}$
	1423	1·08		1773	20·6
	1473	2·12		1823	44·2
	1573	6·1		1873	51·8
	1673	10·8			
		Hydrogen-copper.			
129	498	$0·16\times10^{-8}$			
	568	1·8			
	603	2·4			
	623	3·6			
	668	10·9			
	723	30·5			
	730	34·3			
		Hydrogen Iron (Etched Surface).			
140	518	$2·4\times10^{-6}$			
	686	17·6			
	739	31·2			
	793	45·3			
	894	92·8			
	999	162·0			
	1052	203·0			

equation which was in good agreement with their own measurements for the diffusion of hydrogen through platinum—

$$D=(k/d)\,P^{\frac{1}{2}}T^{\frac{1}{2}}\,e^{-b/T} \quad . \quad . \quad . \quad . \quad . \quad (1)$$

where D is the rate of diffusion per unit area of surface, P is the gas pressure, T the temperature, d the thickness of the metal, and b a constant for a particular gas-metal system. In practice $T^{\frac{1}{2}}$ is negligible compared with the exponential term and may be omitted from the equation.

The effect of temperature on the diffusion constant D (D = volume of gas in c.c. at N.T.P. diffusing per second through 1 square centimetre of surface of 1 mm. thickness) for a number of systems is given in Table I.

If the effect of temperature on the rate of diffusion is represented by the equation

$$D=k'\,e^{-b/T} \quad . \quad . \quad . \quad . \quad . \quad (2)$$

then on plotting log D against 1/T a straight line should be obtained. This was found to be the case with all the systems investigated; b is, of course, given by the slope of the log D—(1/T) graph and is found to be roughly of the same order of magnitude for all the systems. Similarly, the constant k in equation (1) does not vary greatly for various systems.

The dependence of the rate of diffusion on pressure

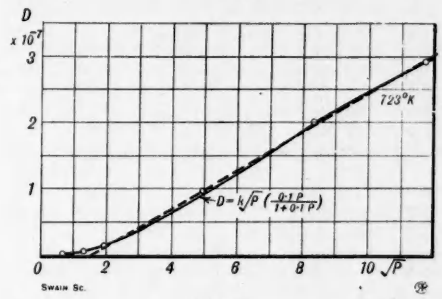


FIG. 1

according to the Richardson equation should be represented by

$$D=k''\,P^{\frac{1}{2}} \quad . \quad . \quad . \quad . \quad . \quad . \quad (3)$$

From measurements made by Sieverts‡ and others it is shown that the solubility of diatomic gases in a number of metals is proportional to the square root of the pressure. This can be easily shown to be in accordance with the hypothesis that the gas dissolves as atoms and not as molecules. As the rate of diffusion will depend on the concentration of gas in the metal, we might expect that D would also be proportional to $P^{\frac{1}{2}}$. In Fig. 1 is shown the plot of D against $P^{\frac{1}{2}}$ for the authors' observations in the system hydrogen-copper, and although at the higher pressures

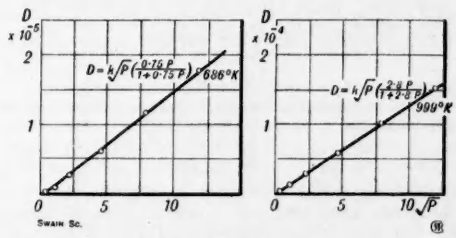


FIG. 2

the points lie very closely on a straight line, that is a definite departure from this relation at low pressures where the rate is less than that required by equation (3). Fig. 2 shows the results for hydrogen and iron

† *Phil. Mag.*, 8, 1904, 1.

‡ *Zeit. Metallkunde*, 1929, 31, 27.

(etched). The results for the effect of pressure on the rate of diffusion are given in Table II.

TABLE II.—*Effect of Pressure on Rate of Diffusion.*

Pressure, mm.	Rate (D).	Pressure, mm.	Rate (D).
Hydrogen Nickel at 1023 Deg. K.			
0·0066	$0·82\times10^{-6}$	4·7	$26·4\times10^{-6}$
0·0205	1·57	9·3	42·8
0·0286	2·26	21·5	63·9
0·0547	3·14	47·0	100·0
0·0950	4·08	105·5	149·0
Hydrogen Copper at 723 Deg. K.			
0·52	$0·017\times10^{-7}$		
1·83	0·054		
3·97	0·147		
24·5	0·96		
69·0	2·03		
129·5	2·93		
Hydrogen Molybdenum at 1673 Deg. K.			
1·68	$2·3\times10^{-5}$	42·5	$12·8\times10^{-5}$
10·0	6·3	69·0	18·2
14·6	8·3	81·0	18·5
25·0	10·2	81·0	19·0
27·0	10·5	95·5	20·4
39·7	13·4	126·0	23·9
Nitrogen Molybdenum at 1773 Deg. K.			
4·4	$0·39\times10^{-6}$		
22·0	1·06		
40·5	1·51		
67·0	1·64		
79·5	2·12		
86·5	2·11		
Hydrogen Iron (Etched Surface).			
At 686 deg. K.		At 1000 deg. K.	
0·094	$0·12\times10^{-6}$	0·094	$1·81\times10^{-6}$
1·37	0·71	1·37	11·9
5·1	2·52	5·1	27·5
21·5	6·11	22·0	59·0
60·7	11·4	61·7	101·5
140·0	17·6	142·0	153·0

Smithells and Ransley point out the probable importance of adsorption in relation to diffusion and indicate that all diffusion must be preceded by adsorption. At low pressures the amount of gas adsorbed is usually only a small fraction of a unimolecular layer, and if diffusion only occurs through the regions of the surface covered by adsorbed gas the effective area is $x\,\theta$, where x is the total area and θ the fraction of the surface covered. They use the well-known Langmuir isotherm,

$$\theta=\frac{a\,P}{1+a\,P},$$

where a is a constant for each system. Equation (3) then becomes

$$D=k''\,P^{\frac{1}{2}}\frac{a\,P}{1+a\,P}\quad\ldots\quad(4)$$

At high pressures the term within the brackets approaches unity and the equation approximates to (3). The authors have determined a from their own experimental values and show that the above equation fits the graph of the observed values quite well. Moreover, values of a determined from adsorption data are in fairly good agreement with the values obtained from diffusion measurements.

We might expect that if adsorption were the controlling factor in diffusion then various surface treatments such as activation or poisoning, which are known to affect adsorption, would influence diffusion rates. It is known that the real surface of an adsorbent may be as much as twenty times the measured surface and should therefore affect the rate of diffusion, particularly at low pressures. This is found to be the case. Rates of diffusion of hydrogen for low pressures of the gas and iron and nickel tubes after various treatments are given in Table III.

TABLE III.—*Effect of Surface Treatment on Rate of Diffusion.*

Metal.	Treatment.	Temp. ° K.	Pressure, mm.	D.
Nickel	Polished	1023	0·042	$1·39\times10^{-6}$
	Oxidised and reduced	1023	0·042	2·70
	Polished	1023	0·091	2·91
	Oxidised and reduced	—	—	4·23
Iron	Polished	673	0·77	$0·47\times10^{-7}$
	Etched	673	0·77	4·4
	Polished	863	0·073	1·28
	Oxidised and reduced (600 deg.)	863	0·073	0·76
	Further reduced at 800 deg.	863	0·073	1·54

Etching the surface in acid increased the rate of diffusion ten times.

The authors' view of the diffusion process is that the gas is adsorbed on the metal immediately followed by or simultaneously with the dissociation of the gas into atoms. They consider that no other explanation of the $\sqrt{P}$ law is at all probable. The specific nature of diffusion they attribute to the view that diffusion can only occur after adsorption and they point out that while nitrogen diffuses readily through iron and molybdenum it does not diffuse through copper. Hydrogen diffuses readily through a number of metals, but not argon or helium. They state that if helium does diffuse through copper, nickel, or molybdenum, the rate is at least 10^5 times slower than for the common gases. Helium is not adsorbed by metals above −74 deg. Cent. and nitrogen is not adsorbed by copper above room temperature. Again, nitrogen diffuses through iron, chromium, and molybdenum, metals with which under suitable conditions it can form nitrides, but not through copper and nickel towards which it is inert. Smithells and Ransley conclude that weak physical adsorption is not sufficient for diffusion to take place, but activated adsorption with dissociation of the adsorbed gas is necessary.

The authors next consider how the atoms or ions adsorbed in the metal surface penetrate the metal, whether the gas diffuses through the metal lattice, the grain boundaries, or the "Smekal cracks." From experiments made with hydrogen on a single crystal tube of iron, where the measured rate of diffusion was the same as with the polycrystalline material, they conclude that the presence or absence of grain boundaries has no effect.

An adsorbed atom on a metal surface will receive energy from the thermal agitation of the neighbouring atoms and may then either evaporate into the gas or pass into the metal. The presence of a foreign atom will cause distortion of the lattice, and if the force binding the gas and metal atoms is great the metal atom will be less firmly bound to its neighbours and the resulting distortion will assist the passage of the gas atom. On this view diffusion results from the formation of unstable gas-metal associations, the gas atom being passed on from one metal atom to another, the rate and direction of diffusion being determined by the concentration of the gas at different points in the metal.

This paper forms a notable contribution to this important and rather difficult subject, and the authors are to be especially congratulated on the excellence of their experimental technique.

A New Method of Extraction of Gases in Metals.

A NOTE communicated to the Académie des Sciences by Messieurs L. Moreau, G. Chaudron, and A. Portevin* describes their preliminary work on a new method of extracting gases from metals.

Up to the present, extraction of gases (whether adsorbed on the surface, in solution, in gaseous inclusions, or chemically combined) has been effected either by extraction *in vacuo* or by chemical attack. The new method consists of making the metal specimen the cathode in an evacuated discharge tube. With a current of 5 milliampères at 150,000 volts, the cathode does not heat up and the extraction can be made in the cold. The discharge tube is connected with a Langmuir pump backed by a mercury pump.

It is supposed that gas molecules at the surface are ionised and carried away in the electric field, renewal of gas at the surface being effected by diffusion from within the metal. If the sheet metal specimen has a thickness of more than 1 mm. it is necessary to proceed by stages with an interval of twenty-four hours between operations. The specimen may, however, be heated—for example, by induction—and then only one operation is required to remove the gases from a metal specimen several millimetres thick.

So far tests have been made on aluminium, calcium, iron and magnesium. Degassing by the new method is much more rapid than by the old, and, moreover, the quantities of gases extracted are quite different, as is shown, for example, by the following experiments made on aluminium :—

A.—Fused in a vacuum of about 0·001 mm. of mercury.
B.—The new method applied to the same specimen after fusion *in vacuo*.

Method.	Volume of gas, c.c. per 100 g.	Percentage composition.			
		CO_2.	N_2.	CO.	H_2.
A	7·3	1	20	22	57
B	120	2·3	6	15·8	75·7

The nitrogen extracted from calcium and from aluminium was found to include that present as nitride.

That there is a selective extraction of the gas is shown by the results, given below, of an experiment made in the cold on a specimen 4 mm. thick :—

Number of extraction.	Volume of gas, c.c. per 100 g.	Percentage composition.			
		CO_2.	N_2.	CO.	H_2.
1st	76	1·6	6	28	64
2nd	32	—	2	15	83
3rd	24	—	—	8·5	91
4th	17·5	—	—	—	100

The authors express the hope that the joint study of this new method and of the usual extraction *in vacuo* will give information about the various states of the different gases, and so throw light on their metallurgical significance.

* *Comptes Rendus*, July 16th, 1935, 201, 212–214.

Although this preliminary note gives only very scanty information about procedure, it contains enough to show that the new method may yield some surprising results ; but its application will also involve a thorough investigation to track down the source of the unexpectedly large volume of gas which it discloses, and to establish beyond doubt that this is all coming from the metal specimen.

The Use of Alkali Metals as Deoxidisers.

THE high affinity for oxygen of the alkali metals has from time to time encouraged the hope that they may be of use as deoxidisers for various metals and their alloys. The possible use of lithium in this connection is described in the pamphlet "Lithium : Theoretical Studies and Practical Applications," by H. Osborg, published by the Electrochemical Society, 1935. This publication surveys the occurrence and production of lithium, its chemistry and physico-chemical properties, its possibilities in metallurgy as a constituent of alloys—a section illustrated by numerous constitutional diagrams—and its prospects as a deoxidising and refining agent for the high melting point metals, such as iron, nickel, copper, and their alloys.

For carrying out lithium treatment for deoxidising copper a 50 : 50 lithium-calcium alloy is recommended. This is said to give better results than either lithium or a lithium-rich copper alloy. The alloy is placed in an inverted cup provided with openings on the sides and on top. This container is fastened to a rod and the cup dipped into the lower portion of the molten metal. After thorough stirring for a few seconds to ensure complete distribution of the lithium alloy, the molten metal is ready for pouring. During the lithium treatment a gentle, or in some cases a vigorous, boiling of the molten metal is observed. Only about 0·025 per cent. of the lithium alloy is required. The resulting copper has a lithium and calcium content below 0·01 per cent. and is specially suitable for purposes requiring high conductivity. It is said to be quite free from oxide and can be annealed in a reducing atmosphere without any bad effects. The evidence given indicates that the product was good, but does not enable the exact influence of the lithium to be ascertained. For example, there is no indication as to the share which the calcium has in the result, though it is stated that the lithium-calcium alloy is more effective than calcium alone, and Masing, Schumacher and others are quoted as having come to the conclusion that lithium is an excellent deoxidiser for copper.

A somewhat similar application of an alkali metal to the deoxidisation of brass is described by B. A. Weimer in the *Du Pont Magazine*, April, 1935. This paper refers to a brittle sodium-zinc alloy of high melting point, containing 98 per cent. zinc and 2 per cent. sodium and made by the Grasselli Chemical Company, Cleveland, Ohio. The alloy is added, just before pouring, to the molten metal, whose composition has been arranged so as to give the correct zinc content when the additional zinc contained in the alloy is included. The added alloy sinks to the bottom and is decomposed into zinc which passes into solution, and sodium which reacts with oxides giving sodium oxide, which boils at 880 deg. Cent. and so gives rise to a stream of vapour which is beneficial in

removing gas and non-metallic inclusions. The soundness and cleanliness of the metal are said to be greatly improved, and the properties of the brass are not harmed by a slight excess of sodium.

The use of sodium-zinc is not limited to the brasses, but it may be used for any copper alloy in which the presence of 2 per cent. of zinc can be tolerated. Thus copper-silicon alloys may be deoxidised by sodium-zinc, and among the resultant advantages are increased fluidity and improved casting properties.

Neither of the publications referred to gives any detailed account of comparative tests in which the deoxidiser used was the only variable. In the absence of a fairly prolonged series of such tests, from which, not only the quality of the metal, but the wear on the crucible and other factors would emerge, it is not possible to decide on the merits of these special methods of deoxidisation as opposed to the more usual practice.

The Influence of Phosphorus on Temper Brittleness.

IN an article in the June number of THE METALLURGIST recent German work on temper brittleness was reviewed and reference was made to the latest theory, enunciated by Bischof, to the effect that temper brittleness was due to heterogeneity of the ground mass of the steel produced by precipitated chromium or manganese derived from the decomposition of a chromium-rich or manganese-rich carbide at a rate faster than these elements could dissolve in the iron. This explanation appeared to many who took part in the discussion of Bischof's paper as being unnecessarily complicated. Bennek, in particular, put forward an instance of temper brittleness which did not appear to be capable of easy explanation on Bischof's theory. Bennek's objections have not been long in taking a more substantial form, for in the September number of *Archiv für das Eisenhüttenwesen* (1935/36, 9, 147) he has published an account of his work on "The Influence of Phosphorus on Temper Brittleness."

The view was expressed in our article in June that the ignoring of early research on the subject was perhaps leading to an unnecessary duplication of experimental work and complication of theoretical treatment, and now we find that Bennek is coming back (though not without considerable opposition) substantially to the opinion, put forward by Andrew and Green* in 1920, that the phosphorus content of a steel is the prime cause of its temper brittleness. This suggestion, however, is not referred to and neither are the objections which were raised to it at the time.

Bennek's experimental work was carried out on twenty-two steels containing carbon 0·27 to 0·35, silicon 0·2 to 0·3, manganese 0·4 to 0·6 (except when purposely raised to 1·0 to 1·7 per cent.), together with, in some instances, nickel up to 3·7, chromium up to 1·5, and molybdenum 0·55 per cent. The phosphorus varied from under 0·01 to 0·13 per cent. The steels were made in a high-frequency furnace, cast into 20 kilo. ingots, 115 mm. in diameter, and forged to bars 12 mm. square. The treatment applied before the first series of tests (Table I) was oil hardening from 820 deg. Cent., tempering for one hour at 650 deg. Cent., and cooling in oil or in the furnace at a rate of 0·5 deg. per minute between 650 deg. and 300 deg. Cent. The 0·01 per cent. phosphorus steels containing 1·5 per cent. of either manganese, nickel, or chromium, showed no trace of temper brittleness. When phosphorus was increased to 0·1 per cent. in the plain carbon steel, the deterioration of impact figure on cooling slowly

TABLE I.—*Temper Brittleness of Steels with Different Phosphorus Contents.*

Steel No.	Chemical composition.								Impact figure in mkg. per cm.².		Change in impact figure through furnace cooling. %
	C, %	Si, %	Mn, %	P, %	S, %	Ni, %	Cr, %	Mo, %	Oil cooled.	Furnace cooled.	
									After tempering.		
1	0·30	0·23	0·50	0·012	<0·01	—	—	—	22·0	20·7	— 5
2	0·28	0·22	1·43	0·018	0·011	—	—	—	20·6	20·4	— 2
3	0·30	0·15	0·51	0·011	<0·01	1·59	—	—	21·3	21·2	— 1
4	0·26	0·13	0·50	0·011	<0·01	—	1·48	—	23·2	24·9	+ 7
5	0·28	0·15	0·61	0·11	<0·01	—	—	—	16·6	15·6	— 6
6	0·27	0·20	1·69	0·12	<0·01	—	—	—	15·9	5·6	—65
7	0·28	0·16	0·52	0·12	<0·01	1·59	—	—	19·0	13·9	—27
8	0·25	0·13	0·52	0·12	<0·01	—	1·50	—	21·2	15·9	—25
9	0·29	0·28	0·50	<0·01	<0·01	3·59	0·87	—	20·5	18·5	—10
10	0·26	0·43	0·64	0·01	*0·1*	3·54	0·90	—	12·0	9·7	—19
11	0·29	0·22	0·48	0·012	<0·01	3·58	0·92	—	18·1	14·0	—23
12	0·28	0·28	0·49	0·036	0·015	3·71	0·89	—	19·0	1·4	—93
13	0·28	0·27	0·52	0·066	0·015	3·67	0·86	—	15·2	1·7	—89
14	0·22	0·42	0·63	0·075	*0·073*	3·64	0·87	—	12·0	2·1	—83
15	0·29	0·36	0·33	0·11	0·016	3·72	0·92	—	11·0	0·8	—93
16	0·29	0·40	0·29	0·11	0·012	3·66	0·89	—	8·0	2·4	—70
17	0·27	0·30	0·60	<0·01	0·014	3·70	0·91	0·55	17·8	15·7	—12
18	0·27	0·33	0·60	0·046	0·016	3·68	0·92	0·55	17·0	6·2	—62
19	0·27	0·30	0·60	0·081	0·014	3·68	0·93	0·56	15·6	6·2	—60
20	0·28	0·31	0·47	0·13	0·015	3·64	0·93	0·57	15·8	1·8	—89
21	0·28	0·19	*0·96*	0·012	0·01	3·60	0·80	—	16·1	7·6	—53
22	0·28	0·16	*1·65*	0·013	<0·01	3·63	0·89	—	12·0	4·6	—62

* *Journal* of the Iron and Steel Institute, 1920 (1), 621, and (2), 219.

after tempering was unimportant. The steels with 1·5 per cent. of manganese, nickel, or chromium, however, became markedly temper brittle when the phosphorus was raised to 0·1 per cent., susceptibility in the presence of manganese being greater than in the presence of a similar amount of nickel or chromium. The nickel-chromium steel with 0·01 per cent. of phosphorus showed only a 10 per cent. fall in impact figure on slow cooling after tempering, but became strongly temper brittle when the phosphorus was raised to 0·036 per cent. The maximum impact figure obtainable by oil cooling after tempering was not markedly reduced until the phosphorus content was considerably higher. An increase in sulphur content reduced the maximum impact figure obtainable, but did not increase susceptibility to temper brittleness (Nos. 10 and 14). The influence of phosphorus was also observed in the nickel-chromium-molybdenum steels (Nos. 17 to 20), but from two to four times as much phosphorus as in nickel-chromium steels was required to produce the same degree of temper brittleness. Raising the manganese content in nickel-chromium steel produced a similar effect to increase in phosphorus (Nos. 9, 21, 22). Table II shows the reduction in susceptibility to temper brittleness produced by prolonged tempering at 650 deg. Cent.

TABLE II.—*Reduction of Temper Brittleness of Nickel-chromium Steels by Prolonged Tempering or by Prolonged Annealing Before Hardening.*

Steel No.	Heat-treatment.	Impact figure in mkg./cm.².		Change in impact figure through furnace cooling. %
		Oil cooled.	Furnace cooled.	
		After tempering.		
13	1 h. 820°/Oil, 1 h. 650° ..	15·1	1·7	—89
	1 h. 820°/Oil, 5 h. 650° ..	13·0	1·2	—91
	1 h. 820°/Oil, 40 h. 650° ..	18·1	2·6	—86
	1 h. 820°/Oil, 80 h. 650° ..	14·6	7·9	—46
	40 h. 1300°/Air,			
	1 h. 820°/Oil, 1 h. 650° ..	18·2	8·8	—52
15	1 h. 820°/Oil, 1 h. 650° ..	11·0	0·8	—93
	1 h. 820°/Oil, 5 h. 650° ..	14·4	0·8	—95
	1 h. 820°/Oil, 40 h. 650° ..	12·8	1·7	—87
	1 h. 820°/Oil, 80 h. 650° ..	16·9	4·1	—76
	1 h. 820°/Oil, 120 h. 650° ..	17·1	9·4	—45
	40 h. 1300°/Air,			
	1 h. 820°/Oil, 1 h. 650° ..	12·0*	9·7*	—19
16	1 h. 820°/Oil, 1 h. 650° ..	8·0	2·4	—70
	1 h. 820°/Oil, 5 h. 650° ..	10·6	3·3	—69
	1 h. 820°/Oil, 40 h. 650° ..	17·8	7·7	—57
	1 h. 820°/Oil, 80 h. 650° ..	17·5	8·3	—53
	1 h. 820°/Oil, 120 h. 650° ..	16·2	8·7	—46
	12 h. 650°/Air,			
	1 h. 820°/Oil, 1 h. 650° ..	12·9	4·4	—66
	12 h. 650°/Air,			
	10 h. 820°/Oil, 1 h. 650° ..	12·5	4·5	—64
	12 h. 650°/Air, 24 h. 820°/Oil 1 h. 650°	12·6	4·4	—65
	120 h. 650°/Air,			
	1 h. 820°/Oil, 1 h. 650° ..	12·9	6·5	—50
	40 h. 1300°/Air,			
	1 h. 820°/Oil, 1 h. 650° ..	12·1	10·5	—13

* Coarse-grained.

Bennek explains the action of phosphorus in the light of the solubility relations of iron phosphide, or other phosphides in α-iron. The solubility of Fe_3P, shown in Fig. 1,† is displaced to a lower concentration of phosphorus by the addition of nickel, and this in a more marked degree in the presence of carbon. The solubility limit is, however, probably always higher than the average phosphorus content of ordinary steels, and Bennek therefore supposes that the effective phosphorus is that concentrated in the segregated regions which formed the interdendritic boundaries of the original cast structure. Evidence is provided that the phosphorus content of the interdendritic regions‡ is higher than the average in the following analytical figures obtained on an average sample of a nickel-chromium steel taken with a drill of 25 mm. diameter, and a sample taken from the interdendritic region with a drill 0·5 mm. in diameter.

	Percentage of			
	P.	Mn.	Ni.	Cr.
Average sample ..	0·009	0·51	3·60	0·86
Interdendritic region	0·023	0·59	3·47	0·92

The improvement in impact figure of the slowly cooled steel with extended time of tempering at 650 deg. Cent., and the fact that the improved susceptibility to temper brittleness is permanent and not reversible, are explained as being due to equalisation of phosphorus concentration by diffusion. If that is the explanation, an even better effect should be produced by a previous annealing at a higher temperature. Such a treatment was found to be very effective (Table II).§ A study of the microstructures of the steels etched to reveal phosphorus distribution gave support to the above interpretation of the effect of long continued heating at 650 deg. Cent.

Bennek endeavours to deduce the effect of the alloy elements on phosphide precipitation from the degree of precipitation hardening which occurs in numerous low-carbon alloys containing about 1·5 per cent. of phosphorus—Fig. 2. These experiments show that phosphide precipitation is increased by nickel, chromium, and manganese, but not by molybdenum and tungsten. Precipitation hardening is strongly increased by the simultaneous presence of nickel and chromium. It is therefore to be concluded that the solubility surface for phosphide is displaced by chromium and manganese to a lower phosphorus content, as Vogel and Baur have already shown it to be displaced by nickel. From this the supposition that temper brittleness is dependent on phosphide precipitation gains support. The addition of molybdenum to the nickel-chromium alloy—Fig. 2—makes very little difference to the Brinell hardness numbers, so the influence of molybdenum and tungsten cannot be explained by their effect on the position of the solubility surface. It has been observed by Houdremont and Schrader that nickel-chromium-molybdenum steels which are non-temper brittle in a single treatment may be made to show temper brittleness by prolonged heating at 500 deg. Cent., and it may thus be conjectured that the influence of molybdenum is to be sought in a diminution of the rate of precipitation rather than in an increase of the solubility limit. To confirm this supposition precipitation hardening experiments were made with carbon-free nickel-chromium and nickel-chromium-molybdenum alloys containing 1·5 per cent. of phosphorus after a preliminary cooling from 1000 deg. Cent. in oil or in the furnace with a rate of cooling between 650 deg. and 300 deg. Cent. of 0·5 deg. per minute. Fig. 3 shows that after oil cooling both alloys, but after slow cooling only the alloy containing molybdenum,

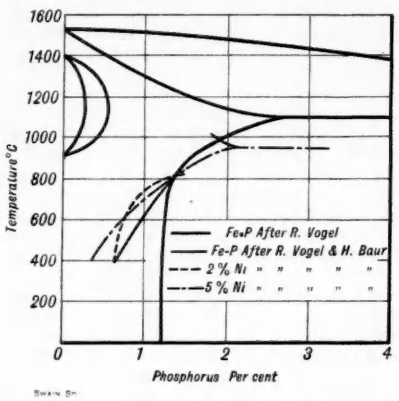
† R. Vogel, *Archiv.*, 1929/30, 3, 369; R. Vogel and H. Baur, *Archiv.*, 1931/32, 5, 269.

‡ In a carbon steel, Northcott (*Journ.*, Iron and Steel Institute, 1934 (1), 153), found 0·10 per cent. of phosphorus in the interdendritic regions, while the dendritic axes contained 0·04 per cent.

§ R. Wijkander (*Jernkont. Ann.*, 1928, 112, 1), previously investigated this treatment.

showed precipitation hardening. The rate of cooling had been slow enough to precipitate all the phosphide from the nickel-chromium steel, but not from the nickel-chromium-molybdenum steel.

The first objection raised in the discussion of Bennek's paper was that urged by Brearley and Monypenny against Andrew's suggestions in 1920, viz., that nickel-chromium and manganese steels of very low phosphorus content sometimes showed



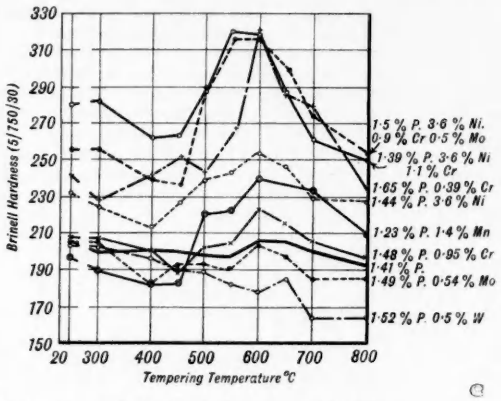
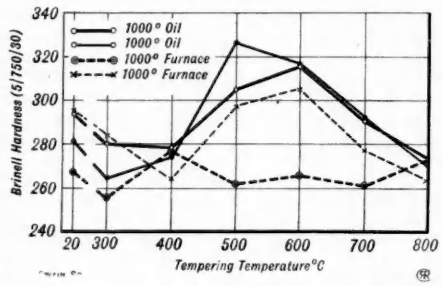

Brinell Hardness (5/750/30)
Tempering Temperature °C
1·5 % P. 3·6 % Ni. 0·9 % Cr 0·5 % Mo
1·39 % P. 3·6 % Ni 1·1 % Cr
1·65 % P. 0·39 % Cr
1·44 % P. 3·6 % Ni
1·23 % P. 1·4 % Mn
1·48 % P. 0·95 % Cr
1·41 % P.
1·49 % P. 0·54 % Mo
1·52 % P. 0·5 % W

FIG. 1—*Solubility of Fe_3P (or of Ni_3P) in α-Iron with 0, 2 and 5 per cent. of Nickel.*

FIG. 2—*Precipitation Hardening of Different Alloys with Less than 0·02 per cent. C, 0·01 per cent. Si and 0·08 per cent. Mn, and with about 1·5 per cent. of phosphorus after Quenching in Water from 1000 deg. Cent.*

marked temper brittleness, and that in such steels it was difficult to suppose that the solubility limit of phosphorus in α-iron had been exceeded even locally. The method of production and degree of purity of the steels had a marked influence on their susceptibility. The case for the "special carbide" theory was put forward and Bennek disclaimed the idea that phosphorus was always exclusively the cause of temper



Composition, per cent.

	C.	Si.	Mn.	P.	Ni.	Cr.	Mo.
Ni-Cr..	0·016	0·01	0·07	1·62	3·79	0·62	—
Ni-Cr-Mo	0·016	0·01	0·06	1·70	3·56	0·62	0·58

FIG. 3—*Influence of Rate of Cooling on the Precipitation Hardening of High Phosphorus Nickel-Chromium and Nickel-Chromium-Molybdenum Steels.*

brittleness. He held that phosphorus in high enough amount, or a corresponding concentration due to enrichment by segregation, will produce temper brittleness by precipitation, but left it an open question how far other precipitated particles, *e.g.*, of carbide, would produce a similar effect. Maurer, on the other hand, declined to regard phosphorus content as a primary cause of temper brittleness, which he ascribed exclusively to the presence of special carbides, though recognising the marked effect that the presence of high phosphorus content might have on the behaviour of these carbides.

Apart from the practical utility of the data presented by Bennek in his paper, perhaps the most interesting point revealed is not so much the dissension between those who believe in the predominating effect of phosphides or of carbides (or it may be added of nitrides or of oxides, though these were not mentioned in this particular discussion), as the general agreement with which temper brittleness is now regarded as a precipitation phenomenon.

Hardening Phenomena in Aluminium Bronze.

ALUMINIUM bronze with up to 12 per cent. aluminium is widely used in the cast and wrought states; according to T. Makuda,[1] the hardness of the alloy can be modified by suitable heat treatment, quenching from above 570 deg. Cent. producing hardening when the aluminium content is low and softening when the aluminium content is high, and a subsequent annealing producing an increase in hardness in both cases. These phenomena are connected with the complex phase changes which occur in alloys with 9·5–16·5 per cent aluminium, the β-phase stable at high temperatures undergoing a eutectoidal decomposition at 570 deg. Cent., the course of which may be modified by variations in the rate of cooling. Quenching and reheating experiments on the eutectoid alloy (12 per cent. aluminium) have indicated that the decomposition of the β-phase is a stepped transformation, the mechanism of which has formed the subject of numerous investigations during recent years.

From electrical measurements and micrographic examination of specimens quenched in water and in toluene, I. Obinata[2] concluded that the decomposition of β takes place in two stages thus: $\beta \rightarrow \beta' \rightarrow \alpha + \gamma$. C. S. Smith and W. E. Lindlieff[3] investigated the

[1] *J.*, Inst. Metals, 1928, 39, 67.
[2] Mem. Ryojun Coll. Eng., 1928, 2, 205; 1931, 3, 87, 285.
[3] "Trans.," Amer. Inst. Min. Met. Eng., Inst. Met. Div., 1933, 104, 69.

change by quenching the β alloys in salt baths maintained at various temperatures between 375 deg. and 565 deg. Cent. and examining the microstructure ; the results confirmed the existence of the β' phase. G. Wassermann[4] studied the lattice structure of the various phases and showed that the β phase at high temperatures has a body centred cubic lattice ($a=2\cdot94$ Å) with random atomic distribution ; on slow cooling the phase decomposes into α, which has a face-centred cubic lattice and γ, which has a γ-brass type of lattice. On rapidly cooling to 300 deg. Cent., however, the cubic body-centred structure is retained, but the atomic distribution becomes regular ; this phase, which is designated β_1, in turn decomposes at lower temperatures into β′, which has a structure similar to α and should therefore be more correctly termed α′. Still further complications are suggested by the work of G. Kurdjumov and T. Steletzky[5], who claim to have discovered a γ′ phase in quenched hypereutectoid alloys; they consider that this is really γ supersaturated with copper.

In view of the many uncertainties still existing in this system, the whole subject has been re-examined by E. Söhnchen,[6] using very pure metals to eliminate possible interference in the reactions by the silicon and iron usually present in ordinary aluminium. By the aid of very sensitive instruments, differential cooling curves were obtained for alloys containing 10·5–13·8 per cent. aluminium, at rates of cooling varying from 5·6 deg. Cent. per minute to 200 deg. Cent. per second. In all cases two heat effects were observed, one at about 500 deg. and the other at about 380 deg. Cent. for the eutectoid alloy. For hypoeutectoid alloys the temperature of these two effects was found to be practically independent of the rate of cooling, but for the eutectoid alloy and for the hypereutectoid alloys the temperature of both effects decreased linearly with increase in the rate of quenching, reaching 435 deg. and 335 deg. Cent. respectively for a rate of 148 deg. Cent. per second. In the hypoeutectoid alloys, a third heat effect (attributed to the formation of γ′) was observed at cooling rates greater than 80 deg. Cent. per second, and the temperature of this effect was also found to be a linear function of the cooling rate, being about 300 deg. Cent. at 80 deg. Cent. per second, and 130 deg. Cent. at 400 deg. Cent. per second. The magnitude of the first heat effect increased, that of the second decreased, and that of the third remained constant with increase in the rate of cooling. The evidence of quenching tests, therefore, confirms the conclusions of the other investigators mentioned above, that decomposition of β takes place in three stages, viz., $\beta \rightarrow \beta_1 \rightarrow \alpha' \rightarrow (\alpha+\gamma)$.

Further support of this theory was obtained by measuring the Brinell hardness of the quenched eutectoidal alloy on heating to 650 deg. Cent. The hardness temperature curve obtained consists of four sections ; up to 300 deg. Cent., the hardness decreased linearly from 140 to 135 ; above 300 deg. there is first a slow then a very rapid fall to 40 at 450 deg.; between 450 deg. and 570 deg. the hardness remains almost constant at 38–40 ; and above 570 deg. Cent. the very soft β is produced. Heating the quenched alloy produces heat effects at 440 deg. Cent. ($\alpha' \rightarrow \beta_1$), and at 525 deg. ($\beta_1 - \beta'$), prolonged annealing at 250 deg. Cent., a slight hardening effect and a slight refinement of the Widmanstätten structure characteristic of the quenched alloys, and annealing at above 350 deg. Cent. a slow decomposition into the ($\alpha+\gamma$) eutectoid. Temperature-resistivity curves obtained on heating the quenched alloys show marked inflexions at 250–300 deg. Cent., and at 400–500 deg. Cent., the second effect being greater and the first less with increasing aluminium content. Electric resistance measurements on quenched chill and sand cast alloys show that the phase changes on heating are more intense with the coarser-grained sand castings than with the relatively fine-grained chill castings.

[4] *Metallwirtschaft*, 1934, 13, 133.

[5] *Ibid.*, 304 ; *cf.* also Kaminsky, Kurdjumov and Neumarx, *ibid.*, 373.

[6] Giesserei, 1935, 22, 289.

Standardisation of the Notched Bar Test Piece.

SINCE the general discussion on the standardisation of the notched bar test piece at Zürich in 1931, much work has been done in Italy and in Germany on the indications of different forms of test piece. There was, it will be remembered, general agreement that the overall dimensions of the test piece should be 10 mm. by 10 mm. by 55 mm., but strong differences of opinion as to the form of the notch, viz., that it should be :

1 mm. radius, 2 mm. deep	Italy
1 mm. radius, 3 mm. deep	Germany
1 mm. radius, 5 mm. deep	France
0·5 mm. radius, 3 mm. deep ..	Germany (subsidiary)
0·25 mm. radius, 2 mm. deep ..	Great Britain

In Italy the foremost investigator of the subject has been Dr. A. Steccanella, who has published three papers on the subject.* The most recent of these expresses his considered conclusions. It deals in the first instance with the three types of test piece having the same notch radius of 1 mm., with different depths of notch. In an elaborate study based on thousands of tests on many different carbon steels (0·10 to 0·67 per cent. carbon) and on nickel, nickel-chromium and nickel-chromium-molybdenum steels, he comes to the important conclusion that no substantial difference exists in the value of the indications of these three forms of test piece, but that when a series of tests favoured one type the degree of preference was always small. Finally, however, Dr. Steccanella sums up in favour of the original Mesnager test piece (notch 2 mm. deep) adopted by the Italians. He does so on the grounds that his tests show that :

(1) The degree of divergence of results on a given material was equal for the two types with notches 2 mm. and 5 mm. deep, and greater for that with notch 3 mm. deep.

(2) In discriminating between the results of different conditions of annealing of the same material, the 2 mm. deep notch was superior to the 3 mm. and 5 mm. notches.

(3) In discriminating between different conditions of heat treatment of the same material, the 2 mm. and 3 mm. deep notches were better than the notch 5 mm. deep.

Finally he concludes : "We thus find that if an international unification is to be arrived at, it is not our country which ought to abandon its own type of test piece, because (it is well to repeat) it is the type on which our greatest experience has been acquired and is, moreover, one which other countries

* *La Metallurgia Italiana*, May, 1932, September, 1933, and February, 1935.

have utilised in parallel with the other types we have considered."

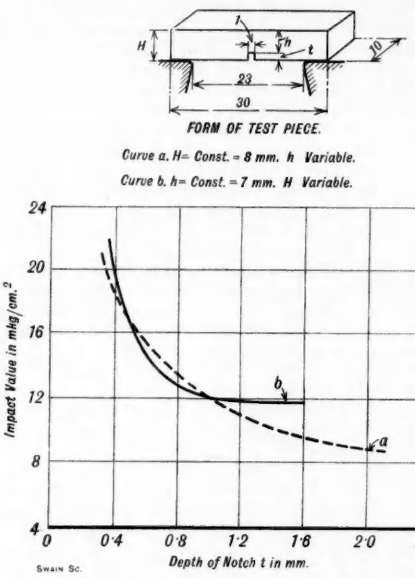
Steccanella has also examined in nearly 1000 tests the indications of the proposed German subsidiary test piece, and definitely disapproves of it. Percentage deviation in results was increased, and, by lowering the maximum figure in tough materials, the numerical difference between tough and notch-brittle conditions was diminished. "It disperses results which it ought to bring together," he says, "and brings together results it ought to disperse." In Italy the need for a test piece other than the Mesnager test piece has never been felt.

Before these conclusions are commented upon, attention may be directed to recent German work. For many years there has been a movement towards the standardisation of the small notched-bar test piece in Germany, and at the Zürich congress, Dr. Moser presented the results of extensive German researches which led to the conclusion that a 1 mm. radius notch, 3 mm. in depth, is most desirable for general purposes, though for testing materials of very great toughness a notch 3 mm. deep with a radius of 0·5 mm. is preferable. After the congress, the German Association for Testing Materials and the Association of German Metallurgists circulated a request for co-operation in comparative tests on different forms of test pieces.

The results of about 9000 such tests on a variety of test pieces (including those examined by Steccanella) have been summarised in a concise form by Dr. R. Mailänder.† The co-operative research has shown that the proposed German standard test pieces yield serviceable indications in practice. They differentiate sufficiently between different types of steel and between steels in conditions of good and bad heat treatment. Deterioration of impact figure by incorrect treatment is sufficiently clearly indicated by the 1 mm. radius notch, but the range of difference between steels which are all naturally tough is widened by the use of the 0·5 mm. radius notch. Mailänder found that, compared with the 5 mm. deep notch of the French test piece, the 3 mm. deep notch has the advantage of greater cross section, and therefore a somewhat greater range of values between tough and brittle material. The 2 mm. deep notch gives, it is true, a still greater numerical difference, but it is held to show certain disadvantages. The somewhat smaller scatter in impact values given by the 2 mm. deep notch in Steccanella's experiments is regarded by Mailänder as accidental and by Steccanella himself as immaterial.

A merit claimed for the 3 mm. deep notch in comparison with the 2 mm. deep is that the notch can readily be prepared in the blank by drilling. Drilling is regarded as essential since grooves of an uncertain degree of sharpness remain at the bottom of a milled notch, while any machine marks left by drilling are circumferential and do not affect the impact figure. The difference between drilled and milled notches may account for a considerable difference in impact figure. For example, in a certain nickel-chromium-tungsten steel, a drilled notch gave an average impact figure of 6·4 mkg. per sq. cm., and a milled notch consistently lower individual results with an average of 5·5 mkg. per sq. cm.

A further argument for the 3 mm. deep notch is based on the variation in energy absorption with increased depth of notch. Reference is made to two series of experiments in which the depth of notch was varied; in one the total thickness of the test piece was unchanged and in the other the thickness below the notch was kept constant. The conclusion was that on introducing a notch there is at first a steep fall in impact figure, but that with increased depth of notch the value tends towards a limit which is reached when the notch attains about one-quarter of the depth of the specimen. To be on the safe side this ratio is regarded as equivalent to a depth of 3 mm. on a test piece 10 mm. thick. The reader of these papers cannot help but feel that the German and Italian authors are magnifying the importance of some of their observations to a degree that is out of all proportion to their true significance. Steccanella admits that the indications of the 2 mm. and the 3 mm. deep notches are for all practical purposes the same, but prefers the 2 mm. deep notch. Mailänder makes an identical admission, but condemns the 2 mm. deep notch. His reasons are, however, not very impressive. It may be agreed that for round notches (though not



Influence of Depth of Notch on the Observed Impact Value of a 5 per cent. Nickel Steel, Tested at 20 deg. Cent.

for sharp notches) milling will introduce serious errors, but there is no difficulty in drilling a 2 mm. diameter hole to form a notch 2 mm. deep, provided that the final operation of machining is to shape down the surface until it is tangential to the drilled hole. The second objection (that a depth of 3 mm. is essential to enable the notch to exert its full and constant effect) is a more substantial one, but unfortunately it is not borne out by the example which Mailänder himself provides. If the diagram which is reproduced above is studied, it will be seen that in the test piece *b* with a constant depth of 7 mm. below the notch, the full effect is produced when the notch is about 1 mm. deep or one-eighth of the thickness of the test piece, while the test piece *a* can never yield a constant limiting value in mkg. per sq. cm. unless the energy absorbed varies as h, which it never does, except in brittle materials. It more frequently varies as some power of h between 1 for brittle materials and

† *Stahl und Eisen*, July 11th and 18th, 1935, 55, 749–754, and 779–786.

2 for very tough materials. Nevertheless, this objection to the 2 mm. deep notch should not be dismissed too lightly ; localisation of deformation by the notch is an important feature of the test, and if a 3 mm. deep notch proved to be more efficient in this respect there would be good reason for using it.

The fact that may give some concern to English users of the test is that neither the German nor Italian investigators have anything to say in favour of the sharp V notch (0·25 mm. radius). The American Society for Testing Materials has now—not without some preliminary investigation—issued a tentative standard specification for notched bar tests, in which the British Standard form of test piece has been adopted.‡

The divergence between the standard practice of Italy, on the one hand, and of England and America on the other, seems to show no sign of lessening, but the gap is bridged by the German subsidiary test piece. The differences between this and the British standard would appear to be of negligible importance when viewed in relation to the empirical character of the test. Yet in the light of the large amount of accumulated experience of a particular form of test piece, the familiar numerical values of its indications and their correlation with composition, heat treatment, other mechanical properties, and practical service, it may be feared that to a demand for international standardisation the reply from all quarters will be, in the words quoted above : " It is not *our* country which ought to abandon its own form of test piece."

‡ " Proc.," Amer. Soc. Testing Materials, 1934, 34, Part I, 1201.

Recent Progress in Alloys for Permanent Magnets.

The Search for Alloys Having Improved Magnetic Properties.

In Report No. 312 of the Constructional Materials Committee of the Verein deutscher Eisenhüttenleute[1] just published, F. Pölzguter reviews the development of improved alloys for permanent magnets from the simple chromium or tungsten steels to the modern complex carbonless iron alloys containing cobalt and nickel with aluminium or titanium.

The best chromium steels for the manufacture of permanent magnets contain carbon 0·9 to 1·1 and chromium 2 to 6 per cent., sometimes with the additions of up to 2 per cent. of tungsten or molybdenum, while the best tungsten steels contain carbon 0·55 to 0·8 and tungsten 5 to 6·5 per cent., the remanence decreasing and the coercivity increasing with increase in carbon content.

The first improvement on these steels was made in 1920, when Honda and Saito[2] introduced steels containing carbon 0·9 to 1·0, chromium 5 to 11, cobalt 5 to 35, and either molybdenum 0·5 to 2 or tungsten 4 to 6 per cent. One of the great advantages of this type of steel is that the cast alloy has practically the same magnetic properties as the hot-forged alloy, and therefore complicated shapes can be cast directly, and require only machining and polishing for finishing.

The next advance was made almost simultaneously in 1931/32 by Rogers and Seljesater[3] in the United States and by Köster[4] in Germany, who found that the presence of carbon in cobalt-tungsten steels was unnecessary for a high coercivity, and that with a suitable adjustment of the cobalt and tungsten contents the heterogeneous structure and state of internal stress necessary for optimum magnetic properties could be obtained by a precipitation-hardening treatment involving quenching from 1250–1300 deg. Cent., and subsequent prolonged annealing at 650–700 deg. Cent. The optimum magnetic properties are obtained with tungsten (or molybdenum) 10 to 20, and cobalt more than 15 per cent. This type of steel has a great advantage over ordinary cobalt steel in that it retains its magnetism up to 650 deg. Cent., whereas plain cobalt steel becomes non-magnetic at little over 300 deg. Cent. and even a moderate increase in temperature above the normal results in a considerable loss of magnetism. Since the cost of magnets of cobalt-tungsten steels is relatively high, and, shortly after their discovery, Mishima[5] introduced his nickel-aluminium-steels, which have even better coercivity combined with good remanence, it is not surprising that relatively little use has been made of them in industry. Mishima's alloys (they are not true steels, in that they are free from carbon) contain nickel 10 to 40, and aluminium 5 to 20 per cent., with small additions of manganese, copper and tungsten, sometimes with part of the nickel replaced by cobalt. At present two types are in general use, viz.:—

(A) Containing nickel 24 to 28, aluminium 12 to 16 per cent., iron remainder.

(B) Containing nickel 24·5 to 30, cobalt 5 to 10, aluminium 9 to 13 per cent., iron remainder.

Still another alloy was introduced by Honda[6] only last year. It contains cobalt 15 to 30, nickel 10 to 25, and titanium 8 to 25 per cent., remainder iron.

The magnetic properties (with a magnetising field strength of 6000 gauss) of these various magnet alloys are summarised in Table I and the (*B.H.*) curves are shown in Fig. 1.

Table I.—*Magnetic Properties of Permanent Magnet Steels.*

Type of steel.	Remanence B_r, gauss.	Coercivity H_c, oersted.	$(B.H.)_{max.}$ mean.
Chromium steel	9,500–11,000	50–75	280,000
6 per cent. W, 0·55 per cent. C	11,000–12,000	55–60	280,000
6 per cent. W, 0·85 per cent. C	9,000– 9,500	70–75	290,000
15 per cent. Co magnet steel	7,500– 8,500	185–195	650,000
36 per cent. Co magnet steel	8,500– 9,500	220–250	950,000
W-Co pptn-hdnd. steel ..	11,000–12,000	100–250	1,000,000
Ni-Al steel A	6,000– 6,300	500–550	1,250,000
Ni-Al-Co steel B	5,800– 6,300	700–750	1,400,000
Co-Ni-Ti steel	6,000– 6,500	750–900	1,650,000

Properties of Aluminium-Nickel-Iron Alloys for Permanent Magnets.

These alloys cannot be worked either hot or cold, nor are they readily machinable, hence magnets can be prepared from them only by direct casting and the castings must be finished by a relatively expensive grinding operation. Chill or sand moulds can be used, but the latter are preferable especially for the more complicated shapes of magnet. Since the alloys have a high shrinkage on solidification, ample

[1] *Stahl u. Eisen*, 1935, 55, 853.
[2] Sci. Rep., Tôhoku Univ., 1920, 5, 417.
[3] " Trans.," Amer. Soc. Steel Treating, 1931, 19, 553.
[4] *Arch. Eisenhüttenwesen*, 1932, 6, 17.
[5] *Ohm*, July, 1932.
[6] Sci. Rep., Tôhoku Univ., 1934, 23, 365.

provision for this must be made in the casting head. For finishing the castings they should be heated to 200 deg. Cent. and ground to the desired shape.

Since the alloys are extremely brittle they can be readily pulverised and the powder mixed with a binder, such as bakelite, for pressing into complicated shapes; in this way excellent magnets may be produced without resorting to grinding, although the magnetic properties of the powder aggregates are inferior to those of the cast alloy, the remanence generally being only half that of castings, but this difficulty can be partly overcome by increasing the cross-sectional area of the magnet.

To obtain the optimum magnetic properties in the cast alloys not only is a careful control of the melting and casting operations necessary, but correct

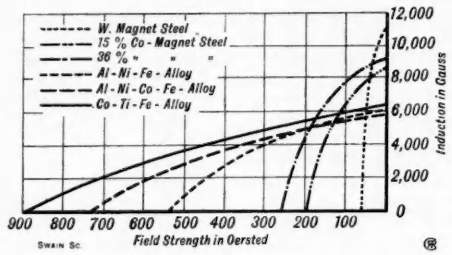


FIG. 1—*(B.H.) Curves for Permanent Magnet Alloys.*

heat treatment subsequently is of paramount importance. Mishima[5] states that annealing at 700 deg. Cent. stabilises the structure and improves the remanence and coercivity; Köster[7] attributes the beneficial effect of annealing to precipitation-hardening, and states that this can also be obtained by suitably regulated cooling of the castings. The best procedure to be adopted depends on the alloy, on the dimensions of the casting, and on the method of casting (sand or chill), but the following examples will give some indication of the results obtainable.

Sand cast bars of various alloys were prepared, 80 mm. long, 30 mm. wide, and 5 mm., 10 mm., 15 mm., 20 mm., and 25 mm. thick, and the (*B.H.*) curves determined. All the alloys contained 24·8 per cent. nickel; for that with 12·7 per cent. aluminium the magnetic properties decreased with increasing thickness of the bar; for that with 14·7 per cent. aluminium the thinnest bar had the worst properties, and there was little difference between the thicker bars; and for that with 15·2 per cent. aluminium, the thickest bar was best and

[7] *Stahl u. Eisen*, 1933, 53, 849.

the thinnest practically useless. A heat treatment involving annealing at 1200 deg. Cent. and slow

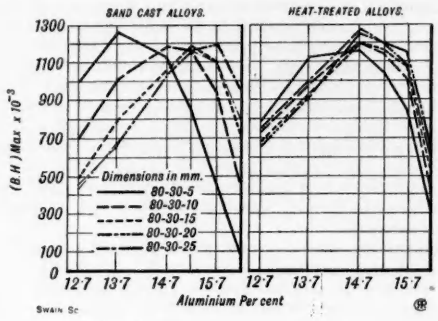


FIG. 2 FIG. 3

Dependence of Magnetic Properties on Aluminium Content and Thickness of Specimen.

cooling to 650 deg. Cent. brought all the bars of the same alloy into practically the same magnetic condi-

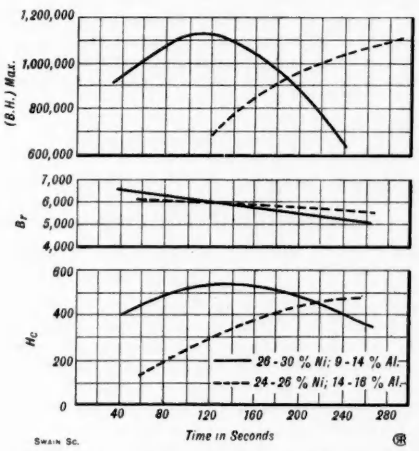


FIG. 4—*Dependence of the Magnetic Properties on the Rate of Cooling of Specimens 80 × 30 × 10 mm.*

tion. Figs. 2 and 3 show the relation between the (*B.H.*) values and the aluminium content for bars of different thicknesses in the cast and heat-treated states respectively; these curves indicate that for a nickel content of 24·8 per cent. the optimum magnetic properties are obtained with 14 to 15 per cent.

TABLE II.—*Influence of Thickness of Bars (40 mm. Long by 30 mm. Wide) of Steel with 28 per Cent. Nickel and 11 per Cent. Aluminium as Cast and Heat Treated.*

	Sand cast.			Quenched from 1,150 deg. Cent.			Quenched and tempered.		
Thickness of bar, mm.	B_r gauss.	H_c oersted.	$(B.H.)_{max.}$	B_r gauss.	H_c oersted.	$(B.H.)_{max.}$	B_r gauss.	H_c oersted.	$(B.H.)_{max.}$
15	3,600	277	323,900	6,000	531	1,156,375	6,275	489	1,197,125
20	3,450	261	270,325	5,775	555	1,338,500	6,050	514	1,204,225
25	2,700	203	173,850	5,850	550	1,159,725	6,150	503	1,220,625

aluminium, the lower value being best for the thinner bars and the higher value for thicker bars.

A second series of alloys with nickel 28 and aluminium 11 per cent. were sand cast into bars 40 mm. long, 30 mm. wide, and 15 mm., 20 mm., and 25 mm. thick. As will be seen from Table II, the magnetic properties in the cast state were very poor. By a subsequent anneal for ten minutes at 1150 deg. Cent., followed by quenching in a 5 per cent. oil-in-water emulsion, properties similar to the best obtainable in the 24·8 per cent. nickel series were obtained, and these could be still further improved by subsequent tempering for forty-five minutes at 660 deg. Cent.

Fig. 4 shows the effect of the rate of cooling from 1200 deg. Cent. on the magnetic properties of the two groups of alloys discussed above ; it will be seen that those alloys with a high nickel and low aluminium content have an optimum rate of cooling, whereas those with a low nickel and high aluminium content develop the best properties when cooled relatively slowly. Nickel-aluminium-cobalt alloys behave similarly to the cobalt-free alloys, except that tempering after quenching is always necessary.

The report concludes with a discussion of magnet shapes for various purposes, and the effect of dimensions on the magnetic flux.

Corrosion of Mild Steels Containing Copper.

Effect of Phosphorus on the Resistance to Atmospheric Corrosion.

The addition of a small amount of copper to mild steel has long been known to improve the resistance of the metal to rusting on exposure to the atmosphere, but the actual behaviour of steels of similar composition, made by different processes, has been found to vary considerably, in particular copper steels made in the open-hearth furnace are frequently much inferior in corrosion resistance to those made in the Thomas converter. The idea that phosphorus is an objectionable constituent of steel is still very widely held, and accelerated corrosion tests in acid solutions have served only to confirm the opinion ; on the other hand, many examples of steel which have withstood atmospheric exposure extraordinarily well and yet have had a high phosphorus content are known. Thus old Roman iron and iron from the Middle Ages have been found to contain an average of 0·15 per cent. phosphorus and the more modern puddled steel 0·09–0·43 per cent.; the best-preserved specimens of these old irons also generally contain 0·11–0·44 per cent. copper. Examination of many structures of mild steel made during the last century and continuously exposed to atmospheric weathering showed that those portions of the structure which were covered with an extraordinarily firm adherent thin film of rust and which had suffered a much smaller loss in weight than neighbouring parts contained a minimum of 0·15 per cent. phosphorus and about 0·2 per cent. copper. These facts suggested to K. Daeves that, despite the unsatisfactory behaviour of high-phosphorus copper steels in acid, a high phosphorus content might be beneficial in steels which are to be exposed to outdoor weathering ; the results of his investigations extending over more than five years have recently been published.*

* *Arch. Eisenhüttenwesen*, 1935–36, 9, 37.

Wires (4 mm. in diameter) of steels containing 0·02 to 0·035 per cent. phosphorus and 0·07 to 1·14 per cent. copper were exposed to the atmosphere near a hydrochloric acid pickling tank and the loss in weight was determined at intervals over a period of sixty-two months ; for comparison, similar wires of steel with 0·06 per cent. phosphorus and 0·33 per cent. copper were also exposed. The loss in weight of the specimens in the first series decreased, first rapidly then more slowly, with increasing copper content, but the comparison wires with a high phosphorus content proved superior even to those with more than 1 per cent. copper and a low phosphorus content. In a rural atmosphere similar results were obtained, although the loss in weight in no case exceeded 7–9 per cent. after exposure for sixty-two months ; at first the plain copper steels behaved similarly to the high-phosphorus steel, but as the film of copper between the rust and the metal became thicker the superiority of the latter steel became evident and increased with time. As the phosphorus content of the steel is increased to 0·3 per cent. the resistance to corrosion of a 0·3 per cent. copper steel rises rapidly, the loss in weight in an industrial atmosphere of the steel with 0·3 per cent. phosphorus being only about half that of the 0·06 per cent. phosphorus steel ; unfortunately, however, steels of very high phosphorus content can be expected to find only a limited use owing to their brittleness unless special procedures are adopted in their manufacture.

Exposure tests over twelve months on a series of steels with various small contents of manganese, copper, tin, and phosphorus showed that the plain tin and copper steels were the most readily corroded and that the smallest loss in weight was obtained with a steel containing manganese 0·74, tin 0·26, copper 0·26, and phosphorus 0·13 per cent., although steel with only a trace of phosphorus and similar tin and copper contents was very little inferior when the manganese content was below 0·25 per cent.

For constructional purposes loss in weight during corrosion is not the most important criterion of the value of steel, since, although different steels may show similar losses in weight after similar exposures, the change in strength or carrying capacity may be very different. Comparative tests in this respect have shown that the loss in strength of steel with 0·3 per cent. copper is about 73 per cent., and that of similar steel with more than 0·1 per cent. phosphorus only about 46 per cent. of that of similar steel with only traces of copper and phosphorus.

Recent work of Kendall and Taylerson† carried out for the American Society for Testing Materials has confirmed the beneficial effects of phosphorus in steels with 0·15 to 0·3 per cent. copper ; their results indicate that the average life of sheets of these steels is doubled by increasing the phosphorus content from 0·01 to 0·10 per cent., and that this effect may be ascribed to the formation of a dark brown, smooth, and very adherent rust film on the high-phosphorus steels.

Corrosion in Various Liquids.

A series of tests on mild steels, containing carbon 0·03 to 0·07, manganese 0·05 to 0·49, phosphorus 0·008 to 0·068, sulphur 0·018 to 0·068, copper 0·07 to 0·64 per cent., and silicon traces only, in the cold-rolled and annealed state, and with and

† " Proc.," Amer. Soc. Testing Materials, 1929, 29, II, 204.

without roughening the surface with emery, has been carried out by F. Eisenkolb.‡ In atmospheric exposure tests the steels with a high copper and with a medium copper and high phosphorus content behaved best, but in 1 per cent. hydrochloric acid the best behaviour was shown by the copper steels containing the minimum of impurities, the high phosphorus steels, irrespective of the copper content, being among the most severely corroded. The presence of inclusions even in the purest steels greatly accelerated corrosion, but retardation occurred in all cases in which a film of copper formed over the surface. In 5 per cent. citric acid and in cold tap water, the losses in weight of the various steels were practically the same, the presence of copper having little effect, but in warm water the copper steels and plain steels with only small quantities of impurities proved superior to the low copper and the more impure steels.

Roughening of the surface had relatively little effect on the corrosion of any of the steels in cold water, and carburising of the surface increased the rate of corrosion in cold water, and in 1 per cent. hydrochloric acid only when the carbon exceeded about 0·15 per cent., but corrosion was severe only when the carbon exceeded about 0·8 per cent.

Since steel structures are usually under some form of tensile stress, it is of interest to determine whether this has any effect on the rate of corrosion in atmospheric exposure or under running water. An investigation on this point has recently been made by E. W. Müller and H. Buchholz,§ who subjected commercially pure iron and six different types of mild steel to the action of the weather and of running water under loads which were (*a*) the equivalent of 90 per cent. of the yield point, and (*b*) sufficient to produce a permanent elongation of 5 per cent. The atmospheric exposure specimens in no case showed any appreciable increase in corrosion over specimens of the same steels which were not under stress, but in running water the stressed specimens all tended to rust more rapidly than the unstressed specimens.

The effect of applied tensile stresses on the scaling of mild steels in air or steam at high temperatures becomes noticeable only when the conditions are such that the scale formed is brittle and tends to flake off; this occurs in most cases at about 600 deg. Cent., but at 700 deg. Cent. or above the scale seems to become plastic and even stresses which cause a slow permanent deformation at this temperature do not increase the tendency to scaling.

Books and Publications.

Molybdenum Steels: Their Manufacture and Application. By J. L. F. VOGEL and W. F. ROWDEN. Published by High Speed Steel Alloys, Ltd., Widnes. 1935. Pp. 103. 5s.

THIS book, written by the authors as a token of the service which High Speed Steel Alloys, Ltd., are prepared, through their technical staff, to render to their customers, may be accepted as a brief but authoritative statement of the properties of molybdenum steels. It describes the commercial forms of the products utilised for the addition of molybdenum to steel, and gives an account of the manufacture, casting, forging, and heat treatment of different classes of molybdenum steel whether of a simple or complex character. The uses and economic value of these steels are also dealt with.

The principles governing the heat treatment of molybdenum steels are enunciated very clearly, and claims are put forward, without exaggeration, for the remarkable influence of even small amounts of molybdenum on the properties of heat-treated constructional steels.

It may be noted in passing that if the recommended method of chemical analysis is exactly followed, the result for a 0·2 per cent. molybdenum steel will depend on the final weighing of only 3 milligrammes of MoO_3. Weighing as lead molybdate is generally preferred. In the method described for the analysis of calcium molybdate no provision is made for the separation of iron, as in that for molyte, possibly owing to the omission of an instruction to filter after precipitation with ammonia.

The beneficial effect of molybdenum does not, of course, extend to all properties of steel and its presence often involves additional care in manufacture. It would be interesting to learn more of the deterioration of impact figure in steel liable to temper brittleness which is said to occur during service at atmospheric temperature. Strain-ageing embrittlement and other forms of embrittlement, as distinct from temper-brittleness, are recognised phenomena, and the authors might have given more details as to the composition and conditions of fabrication and of service of the parts whose impact value, though originally good, is stated to have fallen to a negligible figure in a few months at atmospheric temperature. No evidence is given that a molybdenum steel would be immune from the defect described, nor is any instance known to the reviewer in which a temper-brittle steel properly treated to give a good impact figure has reverted to the notch-brittle condition on simple ageing at atmospheric temperature.

The authors do not minimise the precautions to be taken in the forging, rolling, and stamping of molybdenum steels. Molybdenum does not reduce, but rather increases the need for careful handling. For example, a softening treatment at 600 deg. to 700 deg. Cent. after forging does not by any means render a nickel-chromium-molybdenum steel forging safe for air cooling, except perhaps from quite a low temperature. The authors advise that this soaking should be carried out at 650 deg. to 680 deg. Cent., and should be followed by cooling in the furnace right down to atmospheric temperature. This extreme caution is no doubt justified by experience; at any rate, if it errs, it does so on the safe side. Not the least valuable feature of the authors' practical notes is that they indicate just where this extra caution is required, and where ordinary practice with molybdenum-free steel requires no modification. Some remarks, however, under "Casting" and "Ingot Defects" seem to underestimate the additional care which is required to ensure getting clean and perfectly sound ingots of molybdenum steels.

The commercial applications of all classes of molybdenum steels are described with the aid of some excellent photographic illustrations. The authors and the company which they represent are to be congratulated on having produced so valuable a record of the properties which, in the last ten or fifteen years, have gained for molybdenum a place of great importance among the alloy elements in special steels.

‡ *Korrosion u. Metallschutz*, 1935, 11, 156.
§ *Arch. Eisenhüttenwesen*, 1935–36, 9, 41.